# EFFECTS OF THERMOMECHANICAL PROCESSING ON DAMPING CHARACTERISTICS OF MARTENSITIC Cu - 13.5 w/o Al ALLOY FOR SHIP SILENCING APPLICATION

Edward William Kelly

# NAVAL POSTGRADUATE SCHOOL
## Monterey, California


# THESIS

EFFECTS OF THERMOMECHANICAL PROCESSING
ON DAMPING CHARACTERISTICS OF
MARTENSITIC Cu - 13.5 w/o Al ALLOY

by

Edward William Kelly

December 1977

Thesis Advisor: Jeff Perkins

# REPORT DOCUMENTATION PAGE

READ INSTRUCTIONS BEFORE COMPLETING FORM

1. REPORT NUMBER

2. GOVT ACCESSION NO.

3. RECIPIENT'S CATALOG NUMBER

4. TITLE (and Subtitle)
Effects of Thermomechanical Processing on Damping Characteristics of Martensitic Cu-13.5 w/o Al Alloy for Ship Silencing Application

5. TYPE OF REPORT & PERIOD COVERED
Master's Thesis; December 1977

6. PERFORMING ORG. REPORT NUMBER

7. AUTHOR(s)
Edward William Kelly

8. CONTRACT OR GRANT NUMBER(s)

9. PERFORMING ORGANIZATION NAME AND ADDRESS
Naval Postgraduate School
Monterey, California 93940

10. PROGRAM ELEMENT, PROJECT, TASK AREA & WORK UNIT NUMBERS

11. CONTROLLING OFFICE NAME AND ADDRESS
Naval Postgraduate School
Monterey, California 93940

12. REPORT DATE
December 1977

13. NUMBER OF PAGES
58

14. MONITORING AGENCY NAME & ADDRESS(if different from Controlling Office)

15. SECURITY CLASS. (of this report)
Unclassified

15a. DECLASSIFICATION/DOWNGRADING SCHEDULE

16. DISTRIBUTION STATEMENT (of this Report)
Approved for public release; distribution unlimited.

17. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, if different from Report)

18. SUPPLEMENTARY NOTES

19. KEY WORDS (Continue on reverse side if necessary and identify by block number)
Martensitic transformation
Damping capacity
Ship silencing

20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

Cu-Al alloys demonstrate high damping capacity in the martensitic state. The specific damping capacity in the γ' martensite of a Cu-13.5 w/o Al alloy was found to vary with grain size. Since platelet length increased with grain size while platelet width remained relatively invariant, it can be inferred that martensitic platelet motion was an active mechanism for damping.

(20. ABSTRACT Continued)

Several thermomechanical processes were explored to establish positive grain size control. It was found that grain nucleation and recrystallization and grain growth were very sensitive to the usual parameters of prior strain, strain rate, annealing temperature and annealing time. Severe brittleness, reversible shape memory effects, and pseudo-elasticity were encountered during the process experimentation.

Approved for public release; distribution unlimited

# Effects of Thermomechanical Processing on Damping Characteristics of Martensitic Cu-13.5 w/o Al Alloy for Ship Silencing Application

by

Edward William Kelly
Lieutenant, United States Navy
B.S., United States Naval Academy, 1971

Submitted in partial fulfillment of the
requirements for the degree
of

MASTER OF SCIENCE IN MECHANICAL ENGINEERING

from the

NAVAL POSTGRADUATE SCHOOL
December 1977

Approved for public release; distribution unlimited

# Effects of Thermomechanical Processing on Damping Characteristics of Martensitic Cu-13.5 w/o Al Alloy for Ship Silencing Application

by

Edward William Kelly
Lieutenant, United States Navy
B.S., United States Naval Academy, 1971

Submitted in partial fulfillment of the
requirements for the degree
of

MASTER OF SCIENCE IN MECHANICAL ENGINEERING

from the

NAVAL POSTGRADUATE SCHOOL
December 1977

## ABSTRACT

Cu-Al alloys demonstrate a high damping capacity in the martensitic state. The specific damping capacity in the $\gamma'$ martensite of a Cu-13.5$^{w}$/o Al alloy was found to vary with grain size. Since platelet length increased with grain size while platelet width remained relatively invariant, it can be inferred that martensitic platelet motion was an active mechanism for damping.

Several thermomechanical processes were explored to establish positive grain size control. It was found that grain nucleation and recrystallization and grain growth were very sensitive to the usual parameters of prior strain, strain rate, annealing temperature and annealing time. Severe brittleness, reversible shape memory effects, and pseudoelasticity were encountered during the process experimentation.

TABLE OF CONTENTS


I. INTRODUCTION ---------- 9

A. GENERAL ---------- 9

B. OBJECTIVES ---------- 13

II. EXPERIMENTAL METHODOLOGY ---------- 14

A. MATERIAL SELECTION ---------- 14

B. MELTING AND CASTING ---------- 15

C. METALLOGRAPHY ---------- 18

D. THERMOMECHANICAL PROCESSING FOR GRAIN SIZE CONTROL ---------- 18

E. DAMPING CAPACITY MEASUREMENT ---------- 19

F. GRAIN SIZE DETERMINATION ---------- 20

III. RESULTS AND DISCUSSION ---------- 22

A. PROBLEMS ENCOUNTERED DURING THERMOMECHANICAL PROCESSING ---------- 22

B. COLD WORKING ---------- 23

C. HOT WORKING IN THE $\beta$-REGION ---------- 23

D. WARM WORKING IN THE TWO-PHASE, $\alpha+\gamma_2$, REGION ---------- 25

E. THERMAL CYCLING ---------- 26

F. FINAL THERMOMLCHANICAL PROCESS ---------- 27

G. EFFECT OF GRAIN SIZE ON SPECIFIC DAMPING CAPACITY ---------- 28

IV. CONCLUSIONS ---------- 30

V. RECOMMENDATIONS ---------- 31

VI. APPENDIX A FIGURES ---------- 35

LIST OF REFERENCES ---------- 55

INITIAL DISTRIBUTION LIST ---------- 58

## LIST OF FIGURES


| Figure | | |
|---|---|---|
| 1. | Cu-Al Phase Diagram | 35 |
| 2. | Temperature of Transformation on Cooling as Modified by Aluminum Content [20] | 36 |
| 3. | Furnace and Crucible Used for Casting | 37 |
| 4. | Two-piece Mold Used for Castings | 38 |
| 5. | Furnace and Rolfing Mill Used for Thermomechanical Processing | 39 |
| 6. | Torsional Pendulum | 40 |
| 7. | Damping Specimen Dimensions | 41 |
| 8. | Specific Damping Capacity Measurement Apparatus | 42 |
| 9. | K1, as-cast | 43 |
| 10. | K1C, Homogenized @ 1000°C for One Hour | 43 |
| 11. | K1C1, Cold Worked Martensite with 3.8% Strain | 44 |
| 12. | K1C1A, Cold Worked Martensite after 925°C Anneal | 44 |
| 13. | K1C2A, Microstructure after Hot Working at 935°C | 45 |
| 14. | K4-3A, Microstructure after Hot Working at 960°C | 45 |
| 15. | K1C4A, Microstructure after Hot Working at 850°C | 46 |
| 16. | K2C Microstructure' | 46 |
| 17. | K2C2 Microstructure after Warm Working at 400°C | 47 |
| 18. | K2C23' After 45 Seconds at 835°C | 47 |
| 19. | K2C2 After 75 Seconds at 835°C | 48 |

20. K2C2 After 105 Seconds at 835°C ---------- 48

21. K2C3 Twenty-minute Thermal Cycle ---------- 49

22. K2C4 Sixteen-minute Thermal Cycle ---------- 49

23. K2C5 Ten-minute Thermal Cycle ---------- 50

24. K4, as cast ---------- 51

25. K4-6. After One Flop-Rolling with 10% Deformation at 835°C (50%) ---------- 51

26. K4-6A (h = 0.1385"). Annealed at 835°C for 90 Seconds ---------- 52

27. K4-6B (h = 0.1390"). Annealed at 835°C for 80 Seconds ---------- 52

28. K4-6C (h = 0.1300"). Annealed at 835°C for 70 Seconds ---------- 53

29. K4-6D (h = 0.1050'). Annealed at 835°C for 60 Seconds ---------- 53

30. Specific Damping Capacity Versus Prior Austerite Grain Size ---------- 54

## LIST OF TABLES

Table

I. Typical Properties of Engineering Alloys ---- 32

II. Spectrographic Analysis of Cast Alloys ------ 33

III. Specific Damping Capacity Data for Samples with Different Grain Size ---------- 34

# I. INTRODUCTION

## A. GENERAL

Ship silencing is becoming an increasingly important design requirement in the construction of naval vessels. It has long been a consideration, as the performance of men and equipment is adversely affected by excessive noise and vibration. Today, however, advances in acoustic devices have made ship silencing a critical design requirement. Weapon platforms now must be quiet enough to escape detection by passive listening sonars, as well as quiet enough not to interfere with their own acoustic detection devices.

The United States Navy's traditional approach to ship silencing has been vibration isolation. With this approach, sources of vibration and noise are placed on resilient energy-abosrbing mounts. This affords a reduction in the transmission of noise and vibration from the equipment to the ship's structure, and beyond to the environment. Unfortunately, this method is not completely effective because these mounts are designed for a dual purpose: vibration isolation and shock absorbtion. The damping requirements for the two purposes are dissimilar, and compromise measures have led to the acceptance of less than optimal solutions to both problems. This, coupled with the considerable space and weight requirements for resilient mounts, has

led to the search for new approaches to vibration and noise control.

Recently, the trend in ship silencing has been to design equipment in such a manner as to reduce or eliminate the machine's inherent noise and vibration. This current trend in design can be approached from many directions. Among the most practical are:

1. Accurate balancing of rotating components.
2. Selection of operating frequencies far removed from resonant frequencies.
3. Proper component tolerances.
4. Materials selection.

Of these, proper materials selection is the fathest from reaching its full potential for noise and vibration reduction applications.

The materials commonly used in marine construction have traditionally been selected for strength, fatigue resistance and corrosion resistance. The internal damping capacity of the material has not been a serious consideration in the past. The development of a material that combines a high internal damping capacity (commonly measured as specific damping capacity, SDC, percentage strain energy dissipated per stress cycle) with the necessary strength and resistance properties would make an invaluable contribution to ship silencing efforts.

Several copper-manganese alloys that meet the strength requirements for most shipboard applications and have

high specific damping capacity have been developed and are available commercially. Among these are Sonoston (nominally 37Cu-54.25Mn-4.25Al-3Fe-1.5Ni) [1] and Incramute I (nominally 58Cu-40Mn-2Al) [2]. These and other copper-manganese alloys have ultimate tensile stresses and yield stresses comparable to mild steel, but have a significantly higher damping capacity (See Table I.).

However, in the evaluation of these materials for ship applications, notably propellors, several deficiencies have been discovered. Low corrosion-fatigue resistance and susceptibility to stress-corrosion cracking [2,3], poor corrosion resistance [2,4], and degradation of SCD at temperatures above 100°C [3,4] have hindered acceptance of these materials by the Navy for shipboard utilization. Research is currently being conducted to correct these deficiencies [5].

Other high damping alloy systems with potential for structural applications are nickel-titanium, copper-aluminum, iron-platinum, and copper-zinc [6,7,8]. With the exception of the nickel-titanium system, very little research has been done on the damping characteristics and other metallurgical properties of these alloys. Several NiTi alloys are commercially available, but investigation of other systems has not advanced to the point of commercial availability of these alloys for high damping applications.

Of the many different damping mechanisms that operate in metals, those of most engineering interest are associated

with a non-equilibrium, martensitic transformation product. This high damping martensitic microstructure usually results from rapid cooling of a parent cubic phase. In the Cu-Mn system, the high temperature face-centered cubic (FCC) phase transforms to twinned face-centered tetragonal (FCT) martensite with rapid cooling [9,10,11]. Cu-Al forms three distinct martensites, $\beta'$, $\beta_1'$, and $\gamma'$ (dependent on Al content) when rapidly cooled from a high temperature body-centered cubic $\beta$-phase [12,13,14]. The high-damping-capacity martensite in the Fe-Pt system is produced by quenching from an ordered face-centered cubic parent structure [5]. Considerable research has been done on factors influencing these transformations, which are the same factors which influence damping mechanisms in the various alloys. The SDC of Cu-Mn alloys has been shown to be stress dependent [11,15,16], temperature dependent [11,16,17], and dependent on prior thermomechanical processing [18]. Similar effects have been observed in the Ni-Ti and Cu-Al systems [6,19].

Previous investigations have led to several proposed mechanisms for the internal damping in these materials. For example, the internal friction could be caused by reversible martensitic plate boundary motion or by reversible micro-twin motion within the plate. Also, dislocation motion or reversible transformations could contribute to high damping capacity.

In the present research, in order to investigate the relationship of martensitic structural response to specific

damping capacity, the assumption was made that the internal friction of Cu-Al $\gamma'$ martensitic microstructures is primarily a function of plate length. It was also assumed that plate length is in turn a function of grain size, while platelet width is assumed to be relatively invariant. The validity of these proposals would be shown by a correlation between grain size and SDC, a correlation which would indicate that martensitic plate boundary motion was in fact a primary mechanism of internal damping in this alloy.

## B. OBJECTIVES

The specific objectives of this research have been to:

1. Select an alloy with suitable metallurgical and damping characteristics for potential structural applications.
2. Develop a casting technique for the chosen alloy.
3. Develop thermomechanical processes to establish positive control of prior austenite grain size and a fully martensitic microstructure at room temperature.
4. Examine the effect of grain size on the specific damping capacity of the alloy.

## II. EXPERIMENTAL METHODOLOGY

### A. MATERIAL SELECTION

Several metallurgical characteristics were considered to be important in the selection of a suitable alloy for this project. A binary system with the capacity for transformation to a high-damping, fully-martensitic microstructure was desired. The martensitic start ($M_s$) and martensitic finish ($M_f$) temperatures had to be above room temperature (~20°C) to permit complete transformation in the material and develop maximum specific damping capacity at ambient temperatures. Complete transformation would also give more consistent microstructures (For instance, the size and distribution of secondary phase particles would not have to be controlled.) in order to better isolate the grain size effects. From a practical standpoint, high SDC at room temperature and a potential for shipboard applications were also considered to be requirements.

There were other selection considerations in addition to the metallurgical characteristics discussed above. An alloy with a twinned martensitic structure was considered desirable for two reasons. First, if this examination did not establish a correlation between grain size and SDC, then it could be deduced that twinning motion within the martensitic platelets was the damping mechanism. Second, future efforts could use this work as a foundation

for the examination of the effect of twinning variations on SDC. Finally, an alloy whose damping characteristics had not yet received significant examination was desired.

A Cu-13.5 w/o Al alloy was selected because it met all of the requirements listed above. The Cu-Al system is a relatively simple binary system (See Figure 1.) in which beta (β) near-eutectoid compositions undergo complete martensitic transformation above 20°C [20]. Between compositions of approximately 13.3 w/o Al and 14 w/o Al, the transformation is to the γ', twinned martensitic microstructure (Figure 2). This alloy composition is similar to commercially available bronzes currently used in many shipboard applications. Although the damping characteristics of several Cu-Al based alloys have been investigated (e.g., Cu-Al-Zn alloys by Delaey et. al. [7,8] and Cu-Al-Ni alloys by Kaufman, et. al. [6,19]), there has not been a conclusive determination of the damping mechanism(s) in these alloy systems.

## B. MELTING AND CASTING

A practical objective of this project was to establish a casting method for production of the alloy. Since the selected alloy had been previously studied by Jellison and Klier [20], the first melt, K1, was performed using a procedure similar to theirs. 99.99 Cu and 99.99 Al in the proper proportions (86.50 w/o Cu-13.50 w/o Al) were melted in a graphite crucible at 1100°C (Figure 3 shows the furnace

and crucible.). After sixty minutes the melt was stirred, returned to the furnace for five minutes, then chill cast in a steel mold (7" X 1.5" X 1") cooled by flowing tap water (~20°C). After machining to remove surface irregularities and defects incurred during casting, the finished bar was not long enough to produce damping specimens for the torsional pendulum to be used for SDC measurement. Microscopic examination of this first casting, K1, revealed a coarse two-phase structure.

Concerned with the possibility that the aluminum content would be significantly reduced by oxidation at the high melting temperature, a second casting was performed using a commercial Al-Cu flux, FOSECO COVERAL #79. The Cu and Al were placed in a graphite crucible and covered with a layer of flux. The furnace temperature was again 1100°C, but the flux facilitated melting, and the melt was stirred after only 25 minutes, returned to the furnace for ten minutes, then poured in a new mold (11.5" X 1" X 1") (Figure 11) which was suspended in a large container of water at room temperature (approximately 20°C). Microscopic examination of the second casting, K2, showed a finer grain with virtually complete martensitic transformation, containing only a very fine dispersion of secondary phase particles near the center of the bar.

The bars K1 and K2 were both homogenized for one hour at 1000°C. Samples were cut and sent for chemical and spectographic analysis to Metallurgical Laboratories of

San Francisco, California (Table II). The aluminum content of K1 was found to be significantly below 13.5 w/o, but K2 was within acceptable limits (Table II) and it was decided to use the second casting technique and the new two-piece mold (Figure 4) for all subsequent castings. The cast ingots were cut into samples approximately 1.25" X .5" X .5" to be used for thermomechanical processing trials.

Difficulty was found in achieving grain size refinement with several thermomechanical processes for castings K1 and K2. This prompted an attempt to decrease the initial grain size in the castings by increasing the heat transfer rate during cooling. Accordingly, the cooling water was agitated by two stirring devices to promote convective heat transfer and thereby increase the heat transfer rate. With this additional innovation, the casting procedure for all subsequent melts was identical to that described for K2. A third melt, K3, was contaminated prior to pouring and was discarded. Microscopic examination of a fourth melt, K4, showed that the attempt to increase the cooling rate was unsuccessful; the casting had a two-phase structure, which, while finer than K1, had no martensitic transformation, showing that the cooling rate was slower than that for K2.

It was then decided to retain a two-phase structure as the starting point for final thermomechanical processing of the SDC specimens, for several reasons. There was a tendency for significant grain growth when K1 and K2 were homogenized,

and since the second phase particles in K4 were very fine and uniformly distributed, homogenization at the expense of grain growth was considered unwarranted and unnecessary. In addition, thermomechanical processing trials in the two phase ($\alpha+\gamma_2$) region (See Figure 1.) showed that this was an acceptable initial microstructure. All specimens for torsional pendulum measurement of SDC were machined from bars cut from this K4 casting.

C. METALLOGRAPHY

All metallographic specimens were prepared as follows:

1. Sand on 1/0 grit emergy paper
2. Sand on 3/0 grit emergy paper.
3. Polish on 1 micron alumina wheel.
4. Polish on a 0.05 micron alumina wheel.
5. 15 second etch with ferric chloride etch (96ml ethanol, 5gm FeCl, 2ml HCL).

Photomicrographs were taken with a Bausch and Lomb optical microscope with a Polarois camera attachment.

D. THERMOMECHANICAL PROCESSING FOR GRAIN SIZE CONTROL

The search for a thermomechanical process to control grain size proceeded by trial and error. First, cold working and recrystallization were tried. Then hot working in the single phase $\beta$ region was explored. Warm working in the two phase $\alpha+\gamma_2$, with recrystallization in the $\beta$ region were also attempted. Finally, a sequence of warm working in the $\alpha+\gamma_2$ region, water quenching, and recrystallization in the

β region was tried. The apparatus shown in Figure 5 was utilized in all thermomechanical processing. The heated bars had to be removed from the furnace and exposed to ambient temperatures for short times prior to rolling in the mill. Thus they cooled slightly before rolling and cooled considerably further during rolling in the ambient temperature mill. There were no facilities available for constant temperature hot or warm working. Exactly controllable parameters with this apparatus included initial temperature of the bar, initial microstructure of the bar, amount of deformation per pass on the rolling mill, mill speed (Low-Medium-High), annealing temperature and annealing time.

Using the known effects of strain, strain rate, and temperature on recrystallization and grain growth as guidelines, processing parameters were selected to achieve either enhanced nucleation rate in recrystallization (for small grain size) or rapid grain growth (for large grain size), in order to obtain final samples with a range of prior austenite grain sizes. Various combinations of the controllable parameters were utilized until processes were found which resulted in the desired grain sizes.

## E. DAMPING CAPACITY MEASUREMENT

Specific damping capacity (SDC) was measured using the torsional pendulum shown in Figure 6. This is a relatively simple method to determine SDC, based on measurement of the amplitude of successive torsional vibration in low

frequency free decay. The torsional pendulum used was obtained from NSRDC, Annapolis, Maryland, and was originally built and used by the U. S. Bureau of Mines, Rolla, Missouri. The frequency of operation was 4-5 Hz.

The test specimens were 0.25 inch square bars with a five inch test section machined to 0.200 inch diameter (Figure 7). A strain gauge was mounted at the midpoint of the test section on each specimen to measure the decay of torsional vibration amplitude. The strain gauge output was amplified and recorded on a Honeywell Visicorder. The complete apparatus is shown in Figure 8.

Specific damping capacity (SDC) was calculated from:

$$SDC = \frac{(A_1^2 - A_2^2)}{A_1^2} \times 100$$

where $A_1$ and $A_2$ represent the amplitudes of successive vibration peaks. This is a comonly accepted measurement of damping capacity and can be converted to the familiar logarithmic decrement ($\delta$) by:

$$SDC = (1 - e^{-2\delta}) \times 100 \qquad [21]$$

The initial angular displacement, corresponding to the initial stress, was constant for all damping tests.

F. GRAIN SIZE DETERMINATION

Grain size during processing trials was determined from optical micrographs of sections cut from the test bars.

Measurement of grain size in the torsional prndulum test specimens was made using the stereographic technique outlined by Guy and Hren [22]. The average grain size, $\bar{L}_3$, was determined from:

$$\bar{L}_3 = \frac{1}{N} \sum_{i=1}^{N} (L_3)_i$$

where N is the number of grain boundary intercepts and $L_3$ is the test line segment length. The optical micrographs used in these measurements were taken from sections cut from both ends of the specimen bars prior to machining. In consideration of the large grain size in specimen #3, the sections were cut longitudinally. In all cases the sections were cut far enough from the bar ends that any end effects from the thermomechanical processing could be neglected.

## III. RESULTS AND DISCUSSION

### A. PROBLEMS ENCOUNTERED DURING THERMOMECHANICAL PROCESSING

In the efforts to develop thermomechanical processes to achieve positive grain size control, several problems were encountered. Nucleation and grain growth were found to be extremely sensitive to the usual factors of prior strain, strain rate, annealing time, and annealing temperature. As would be expected, grain growth in the high temperature (>800°C) β-region was very rapid, making size refinement difficult to achieve. In addition to sensitivity to the normal parameters, this alloy displayed other effects which hampered processing. The material was very brittle and regardless of strain rate or amount of deformation per pass in the rolling mill, the maximum strain that could be achieved during cold or warm working was only about 5% before cracking occurred (Kaufman, et al., reported similar brittleness in an 83 w/o Cu - 14 w/o Al - 3 w/o Ni alloy [19].). Other unique effects, resembling the pseudoelastic effect, and the reversible (two-way) shape memory effect [23], were also observed but not examined since they were not germane to the objectives of this research.

Each type of thermomechanical processing explored and the respective degrees of success in achieving grain size control are described in the succeeding sections.

B. COLD WORKING

The first processing technique attempted was to cold work the material in the martensitic state so as to promote strain-induced grain nucleation and recrystallization when the specimen was annealed. The as cast bar, K1 (Figure 9), was homogenized for one hour at 1000°C and quenched. The resulting martensitic microstructure, K1C (Figure 10), was the starting point for the cold work trials. Several sample bars (approximately 0.5" X 0.5" X 1.25") were repeatedly flop-rolled in the mill, decreasing the roller bite after each flop-roll. Trials were made using bite increments of 0.010", 0.005", and 0.002". The maximum strain achieved before cracking was only 3.8% (Figure 11 shows the strained microstructure.). This was insufficient to induce nucleation and recrystallization when a piece of that bar was annealed at 925°C for 60 seconds. Figure 12 shows that the strain in the microstructure was recovered during the anneal but shows no evidence of nucleation or recrystallization. No further attempts at cold working in the martensitic state were made due to the brittleness encountered in this material at room temperature.

C. HOT WORKING IN THE β-REGION

The next process explored was hot working in the single-phase β-region. This was an obvious technique to promote grain growth, but the possibility of control of recrystallization by the use of high strain rates and low β-region

temperatures was also examined. For this process, repeated flop-rolling was again employed; however, this time the bars were preheated into the β-region, reheated between each flop-roll, and quenched after a final reheat. The starting microstructure was again K1C (Figure 10), with initial bar temperatures of 935°C, 850°C, and 830°C. No brittleness problems were encountered, and a strain of 22% was achieved in the 935°C bar after 12 passes with 0.005" increments and low roller speed.

As expected, in the 952°C bar rapid grain growth occurred, with only two grains in the bar cross-section (Figure 13). Bar S1, the large grain damping specimen, was processed using this technique at 960°C, 0.002" increments and low roller speed. A representative microstructure is shown in Figure 14. Considerably less growth was observed in the 850°C bar after 4 passes with 0.002" increment and high roller speed (Figure 15). These results support the theory that at high strain rate (i.e. high roller speed and large deformation per pass) and low temperature, grain refinement can be achieved. Accordingly, three bars were worked at 830°C; one had two passes with 0.070" increment, the second had 1 pass at 0.080" increment, and the third was rolled in one direction only at 0.250". All were found to be two-phase microstructures after the final reheat and quench. After annealing and quenching to obtain a martensitic structure, some recrystallization and grain refinement were achieved. Initially, this technique was not considered

adequate to warrant further experimentation. However, after the techniques of warm working in the two-phase $\alpha+\gamma_2$ region and thermal cycling were not successful, the thermomechanical process used for grain refinement was developed from this technique.

## D. WARM WORKING IN THE TWO-PHASE, $\alpha+\gamma_2$, REGION

For this process, the sample bars were homogenized in the $\beta$-region at 835°C for 20 minutes, then furnace-cooled to 400°C in the $\alpha+\gamma_2$ region. They were then flop-rolled with a 3 minute reheat at 400°C between passes. Most of the trials were performed with bars (approximate dimensions: 1.25" X 0.30" X 0.30") cut from the second casting after it was homogenized at 1000°C for one hour (K2C). The deformation on the first pass was on the order of 0.012" with three additional passes at 0.003" increments. Figures 16 thru 20 show the microstructural results of a typical trial series using this technique. Figure 16 is the initial microstructure, before heating to 835°C and furnace cooling. After warm working, the bar, K2C2, had a two-phase structure with a fine dispersion of second phase particles (Figure 17). The bar was then cut in three sections which were annealed at 835°C for 45, 75 and 105 seconds and then quenched. After 45 seconds the microstructure was still two-phase, with evidence of the ordering reaction that occurs prior to the $\beta$-phase transformation (Figure 18). The microstructure was fully martensitic after 75 seconds at 835°C and quenching,

with significant recrystallization and β-phase grain reginement achieved (Figure 19). Figure 20 shows, that for the longer annealing time of 105 seconds, rapid grain growth occurs at the 835°C temperature.

During these trials, a brittleness problem was again encountered. Strains achieved in different bars during rolling varied from 2.0% to 5.0% before cracking occurred. At the annealing temperature (835°C), variations in prior strain had a dramatic effect on the annealing times required for grain refinement, with differences of up to 15 seconds for a given amount of reactions in the same size piece.

In spite of these problems, this was considered to be the best process for achieving grain size refinement. Before processing the damping specimen bars, two bars from the fourth melt were processed to try to perfect the technique. These bars were approximately the same cross sectional area as the K2C bars but were longer (3.00" X 0.30" X 0.30" approximately). When these cracked at strains of 1.3% and 2.1%, it was decided to go back to the 830°C hot-working process for grain refinement, in order to avoid the brittleness problem.

## E. THERMAL CYCLING

While developing the $\alpha+\gamma_2$ two-phase warm working process, a thermal cycling process to achieve grain refinement was also explored. Bars from K2C (Figure 16) were cycled between furnaces at 835°C and 400°C, on the premise that the resulting

thermal stresses and phase changes might lead to β-phase grain refinement. Cycles of 30 minutes (ten in each furnace, 16 minutes and 10 minutes were used. There was little, if any, reduction in β-phase grain size with the 20 minute cycle (Figure 21). The 16 minute cycle (Figure 22) and the 10 minute cycle (Figure 23) did show some grain refinement, but the difficulty in establishing correct annealing times prompted the discontinuation of these thermal cycling efforts.

## F. ANALYTICAL DEVELOPMENT OF A FINAL THERMOMECHANICAL PROCESS

After various attempts to empirically discover a simple thermomechanical process which would yield a refined β-phase grain size were unsuccessful, it was obvious that more sophisticated analysis would be required. On this basis, a process which gave satisfactory results, K4-6 (Fgures 24 thru 29), was developed after the thermomechanical processing scheme was analyzed as a transient heat transfer problem. The thermomechanical process employed was intermittently hot working in the β-phase region.

The final thermomechanical process for β-phase grain refinement was to:

1. Preheat the specimen bars to 835°C for an analytically determined annealing time of 165 seconds.

2. Flop-roll once with 0.025" bite (about 8% deformation) and reheat at 835°C for 165 seconds.

3. Make additional rolling passes with 0.002" increments

and reheats as necessary (with microstructure monitored by periodic quenching after reheat and examining sections cut from the bar).

4. Anneal 165 seconds and water quench.

When this process was applied to the longer damping specimen bars (approximately 7.5"), the "reversible shape memory" effect was observed. On the first rolling, the specimen bars bent significantly. This necessitated an adjustment in the reheat and annealing times since the entire bar was no longer in contact with the furnace wall, and the area for conductive heat transfer was reduced. While the bars were "soft" during reheat, they were mechanically straightened in the furnace. When quenched, the bars recurled, making additional anneals and straightening necessary before the bars were straight enough for machining after the final quenching.

Using this process, the smallest grain size achieved in the sample bars was approximately 500μ. In the longer specimen bars, the smallest grain size was 1000μ, but the bar was too deformed for machining. When re-annealed and mechanically straightened, grain growth occurred. The final grain sizes in the damping speciments were 1400μ, 3400μ and 6800μ.

## G. EFFECT OF GRAIN SIZE ON SPECIFIC DAMPING CAPACITY

The torsional pendulum apparatus was used to measure the specific damping capacities of the three damping

specimens. The results are shown in Table III. A simple least-squares curve fit was employed in the graphical presentation of results (Figure 30).

## IV. CONCLUSIONS

1. A Cu - 13.5 w/o Al alloy demonstrates high damping capacity (50%) in the $\gamma'$ microstructure, which could possibly be improved with optimization.

2. SDC varies with grain size, implying that martensitic plate boundary motion is one of the prime damping mechanisms in $\gamma'$ martensite.

3. Grain nucleation and growth in this system are very sensitive to the parameters of prior strain, strain rate, annealing temperature and annealing time.

4. The alloy is very brittle except at high temperatures ($>.7T_m$).

5. This alloy exhibits characteristics similar to those described as reversible shape memory and pseudo-elasticity in other high damping martensitic alloys.

## V. RECOMMENDATIONS

1. Further examination of this alloy should be made to obtain more data points. It should be determined quantitatively whether martensite plate width is invariant with grain size, which would more conclusively show that plate boundary motion is the primary damping mechanism in this system.

2. An experimental procedure to isolate and examine the effect of twin boundary motion on SDC should be developed.

3. The other metallurgical properties of this alloy should be examined to determine if attempts to solve brittleness problems are warranted.

TABLE I.

Typical Properties of Engineering Alloys

| METAL | SDC (% at 5000 psi) | UTS (ksi) | YS (ksi) |
|---|---|---|---|
| Sonoston[1] | 30 | 85 | 40 |
| Incramute[2,3] | 40 | 85 | 45 |
| 55-Nitinol[2] | 40 | 125 | 25[4] |
| Mild Steel[2] | 1.5 | 70 | 50 |
| Brass[2,5] | 0.2 | 47 | 15 |
| Gray Cast Iron[6] | 7 | 20 | -- |
| Cu-13.5w/oAl | 50.4[7] | -- | -- |

[1] [1]

[2] [24]

[3] in optimum damping condition.

[4] exact values dependent on test temperatures relative to transition temperatures.

[5] 70-30 (cartridge) brass, annealed.

[6] as cast [25]

[7] maximum achieved. Stree level undetermined.

TABLE II.

Spectrographic Analysis of Cast Alloys

| ELEMENT | SAMPLE | |
|---|---|---|
| | K1C | K2C |
| Copper* | 87.98% | 86.23% |
| Aluminum | 11.87% | 13.59% |
| Balance | 0.15 | 0.18 |

*Chemical Determination

TABLE III.

Specific Damping Capacity Data for Samples with Different Grain Size

| Specimen Number | Grain Size (μ) | SDC (%) |
|---|---|---|
| S1 | 6800 | 50.4 |
| S2 | 3400 | 40.3 |
| S3 | 1400 | 38.9 |

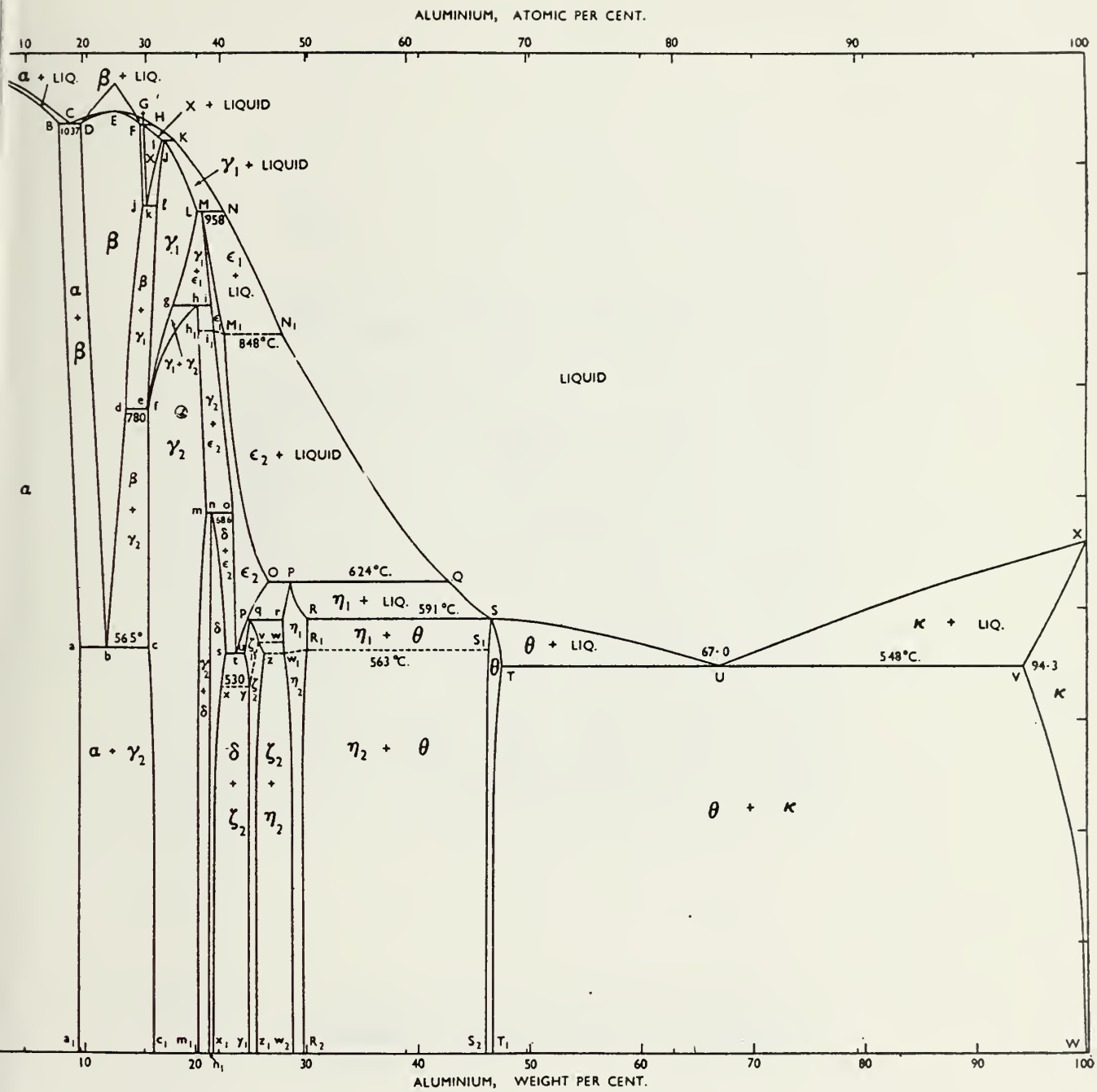


FIGURE 1. Cu-Al Phase Diagram

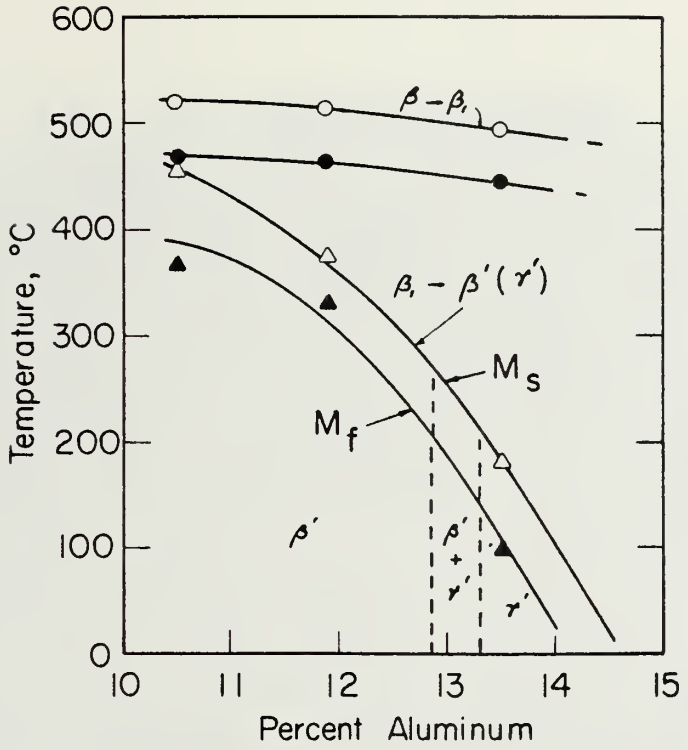


FIGURE 2. Temperature of Transformations on Cooling as Modified by Aluminum Content.[20]

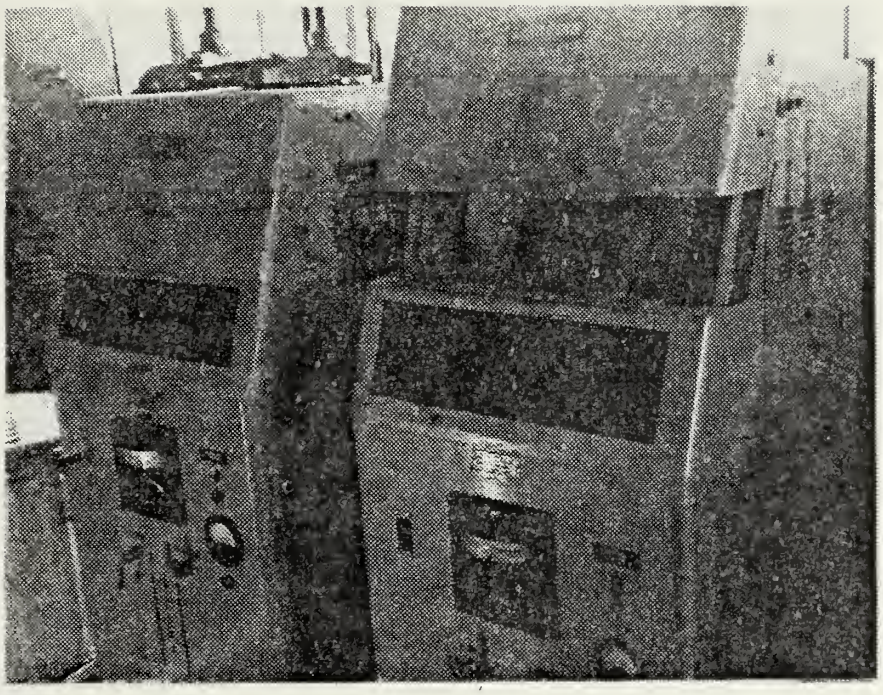

FIGURE 3. Furnace and Crucible Used in Casting

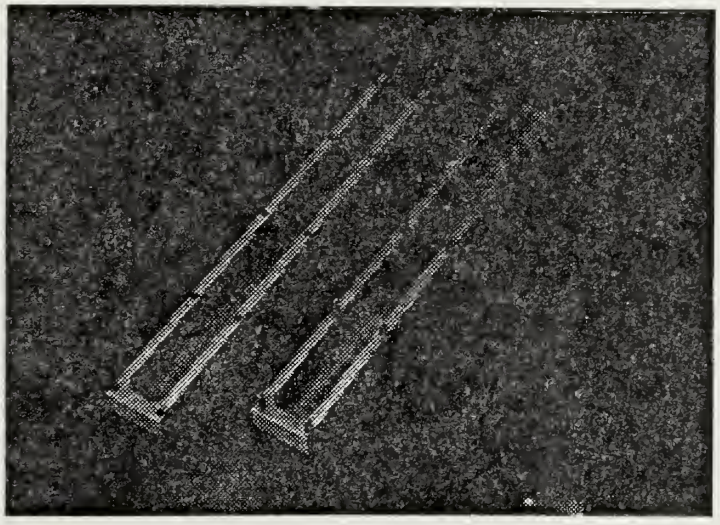

FIGURE 4. Two-Piece Mold Used for Casting

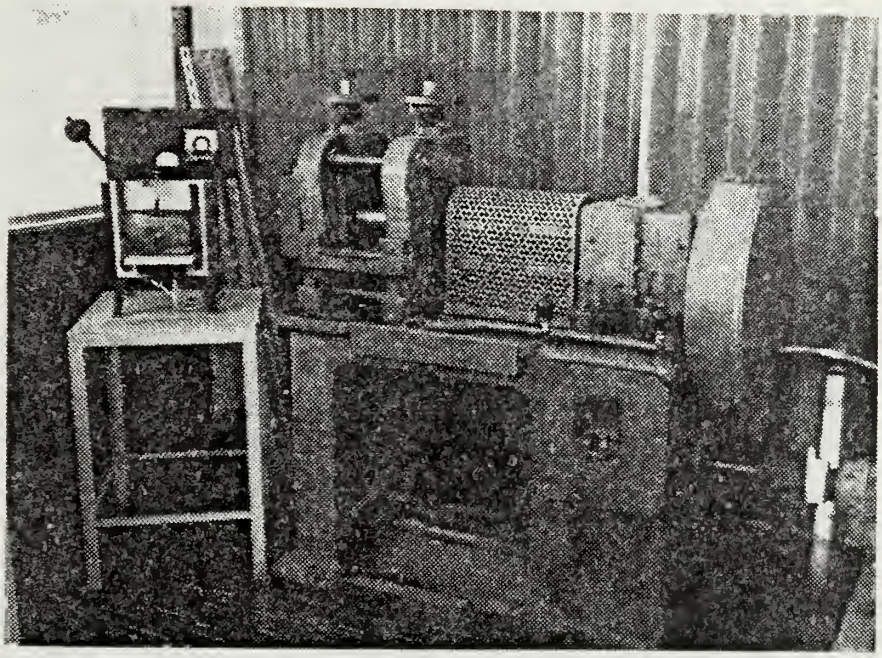

FIGURE 5. Furnace and Rolling Mill Used for Thermomechanical Processing

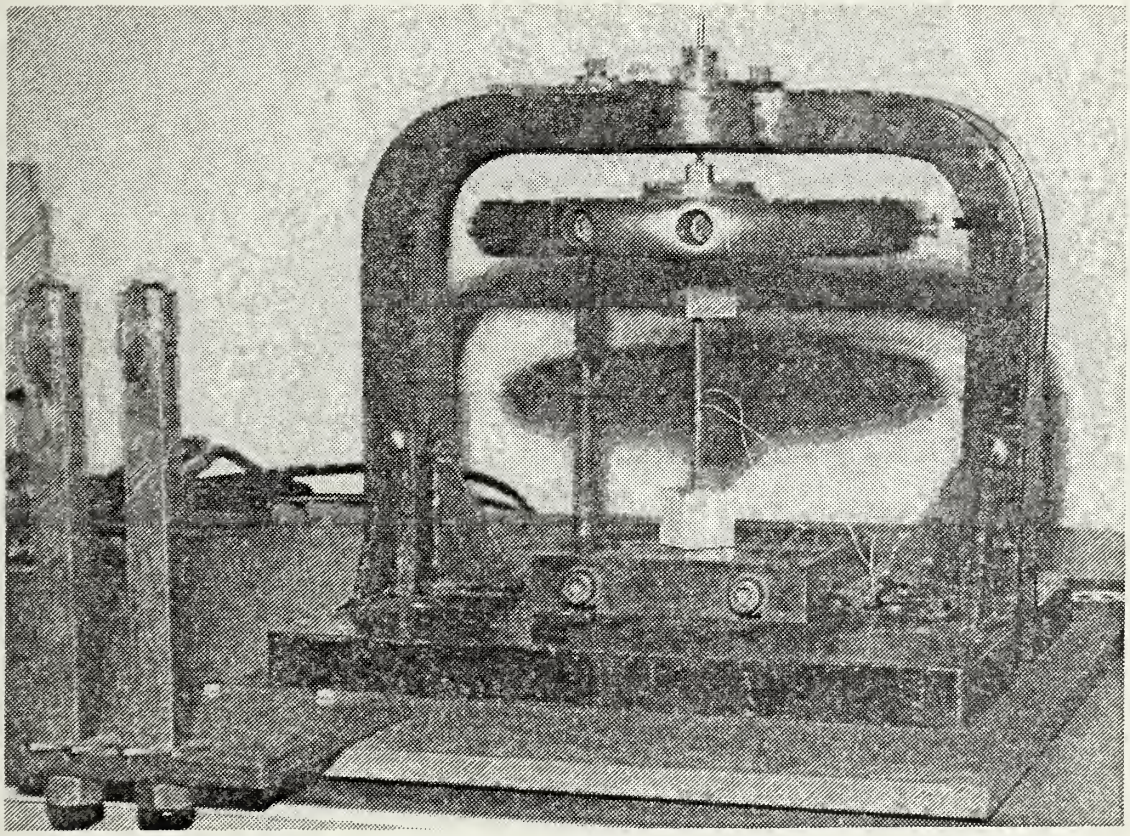

FIGURE 6. Torsional Pendulum

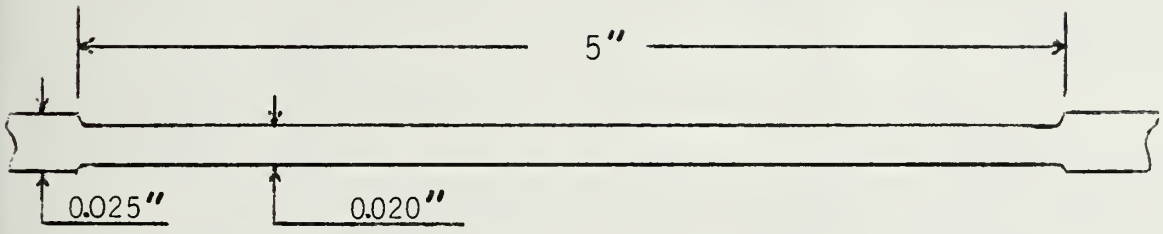


FIGURE 7. Damping Specimen Dimensions

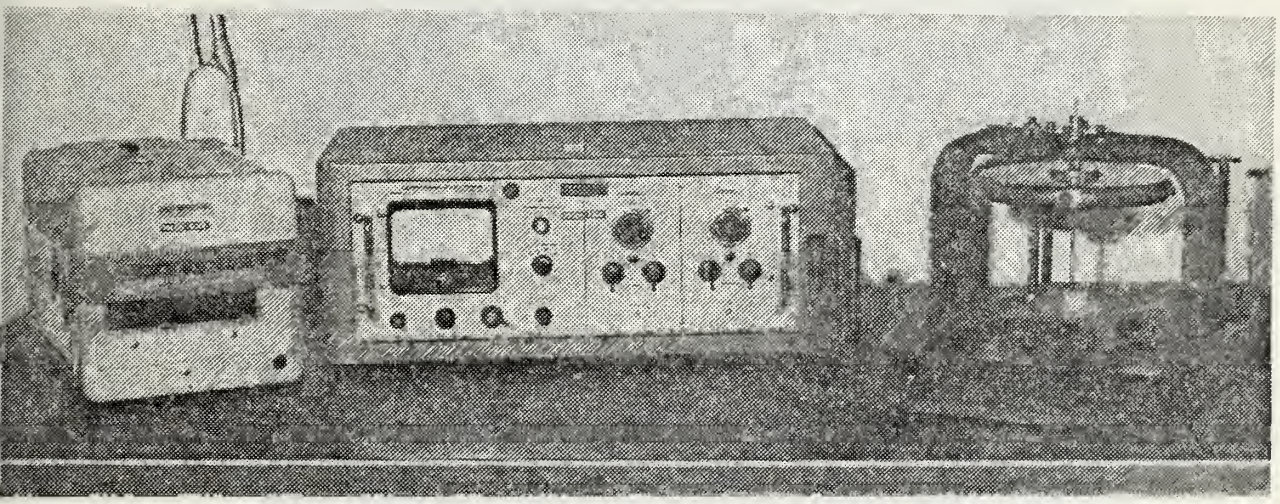

FIGURE 8. Specific Damping Capacity Measurement Apparatus

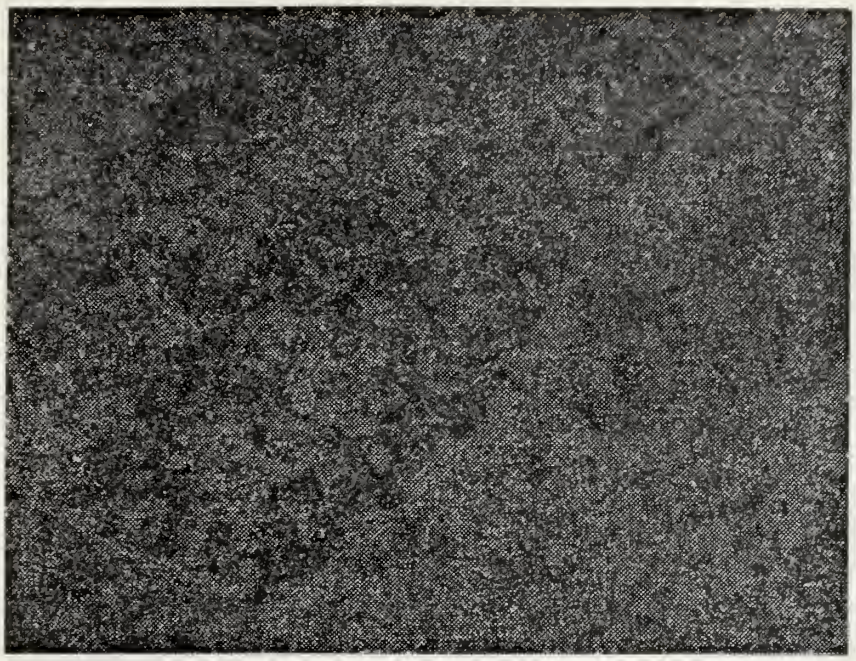

FIGURE 9. K1, as cast (50X)

FIGURE 10. K1C, Homogenized @ 1000°C for One Hour (50X)

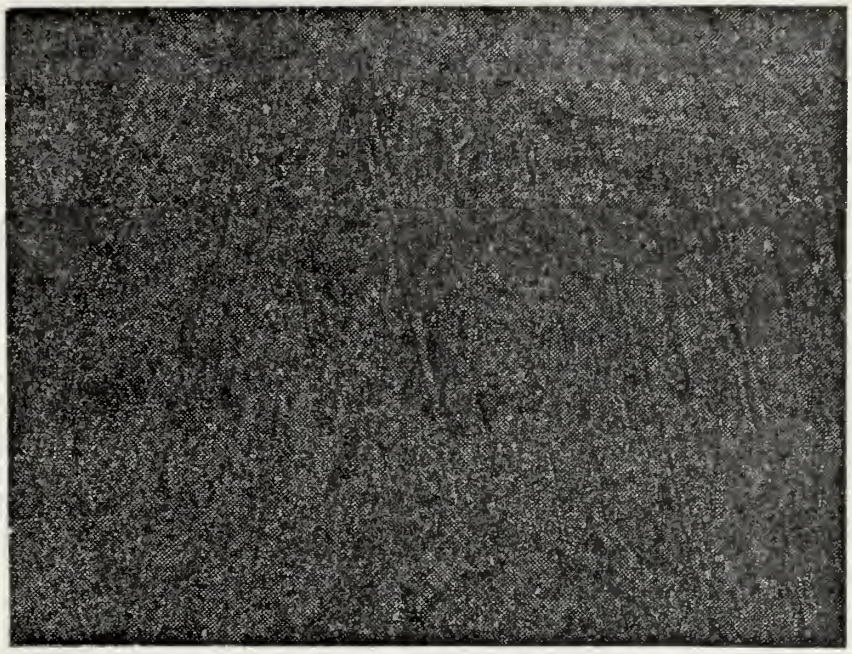

FIGURE 11. K1C1, Cold-Worked Martensite with 3.8% Strain (50X)

FIGURE 12. K1C1A Cold-Worked Martensite after 925°C Anneal (50X)

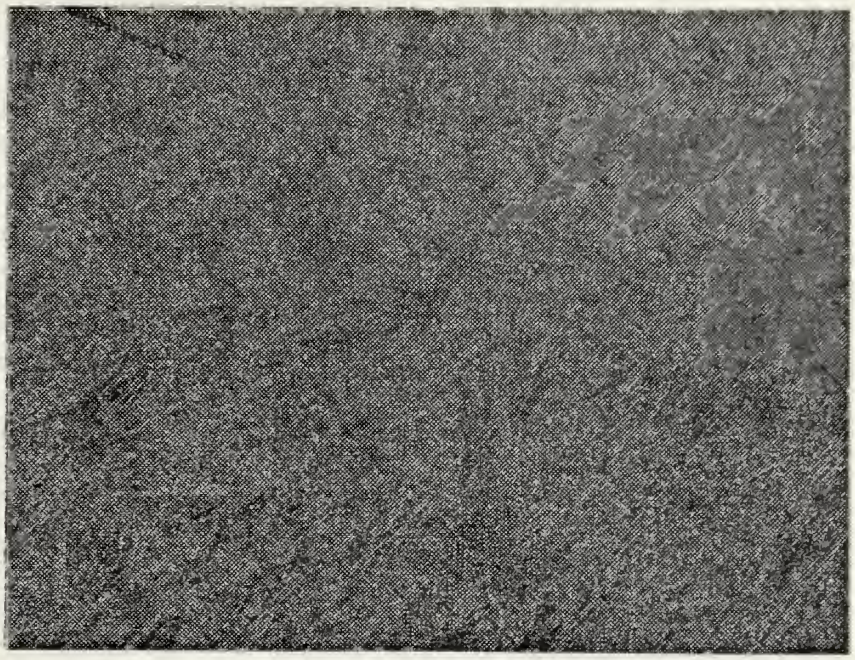

FIGURE 13. K1C2A, Microstructure after Hot Working at 935°C (50X)

FIGURE 14. K4-3A, Microstructure after Hot Working at 960°C (grain size ~ 6800μ) (50X)

FIGURE 15. K1C4A, Microstructure after Hot Working at 850°C (50X)

FIGURE 16. K2C, Microstructure

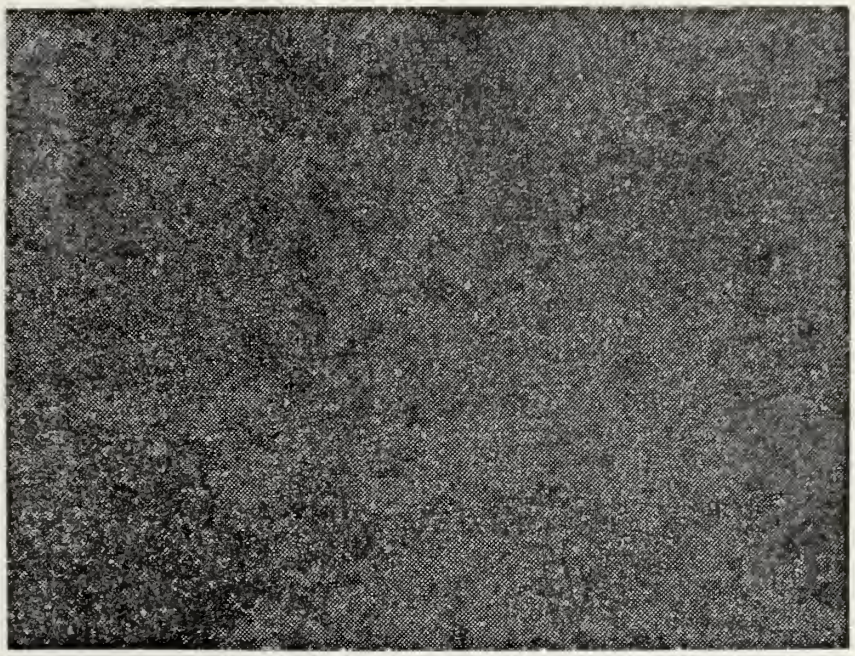

FIGURE 17. K2C2, Microstructure after Warm Working at 400°C (50X)

FIGURE 18. K2C2-3 After 45 Seconds at 835°C (50X)

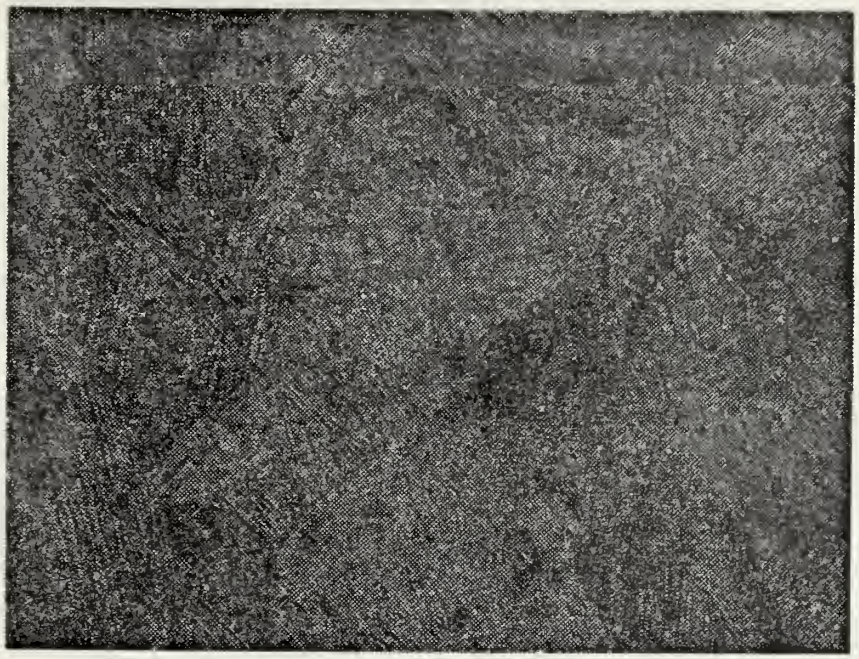

FIGURE 19. K2C2 After 75 Seconds at 835°C (50X)

FIGURE 20. K2C2 After 105 Seconds at 835°C (50X)

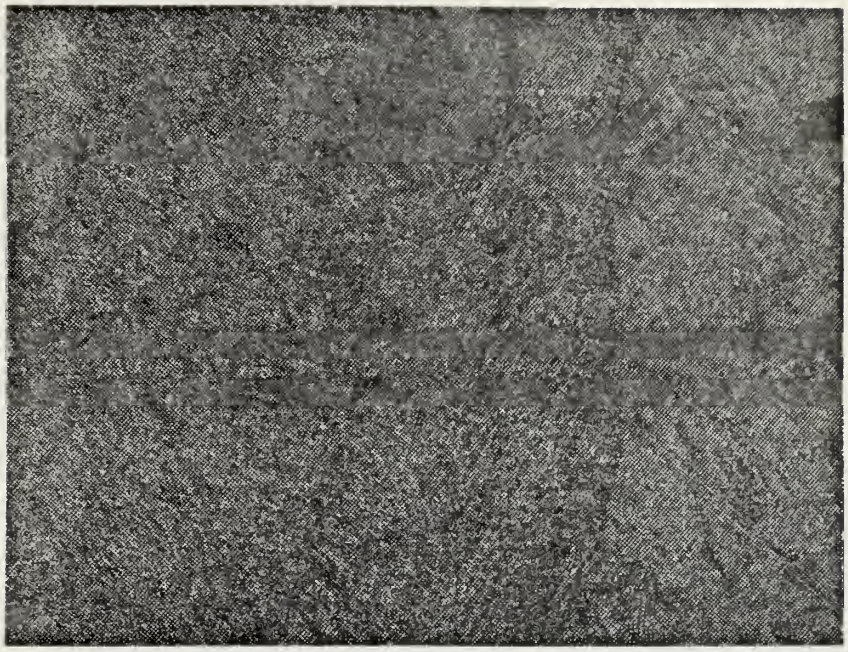

FIGURE 21. K2C3, Twenty-minute Thermal Cycle (50X)

FIGURE 22. K2C4, Sixteen-minute Thermal Cycle (50X)

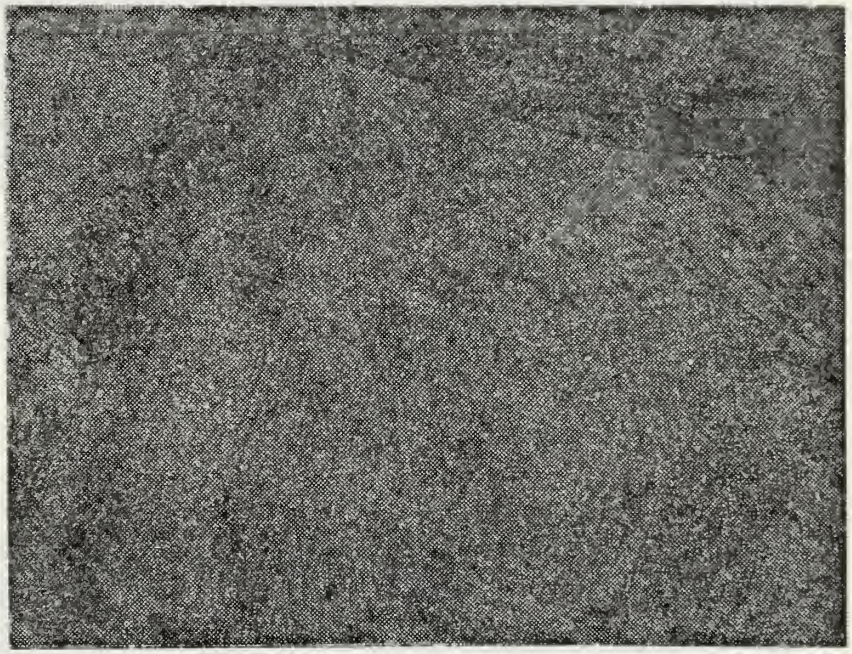

FIGURE 23. K2C5, Ten-minute Thermal Cycle (50X)

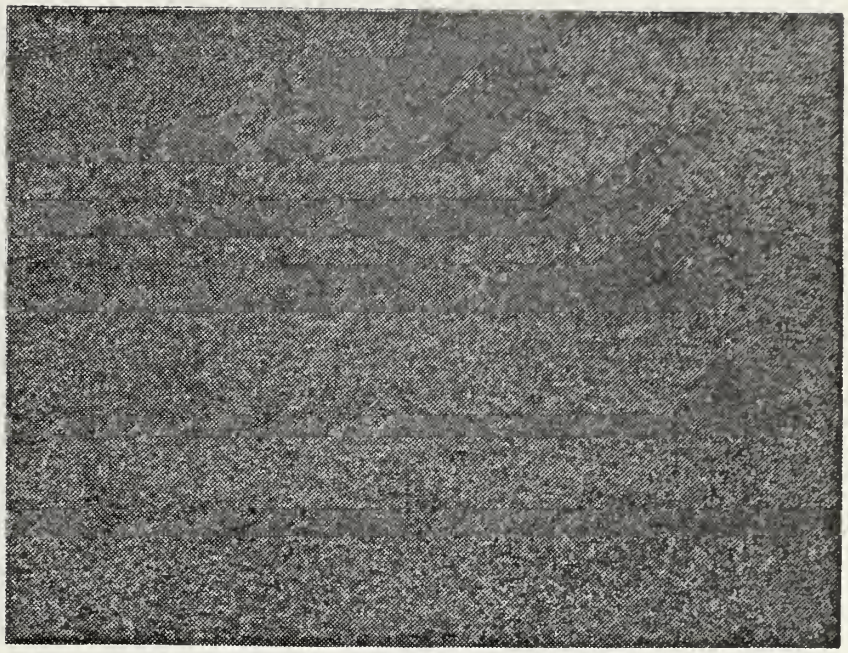

FIGURE 24. K4, As Cast (50X)

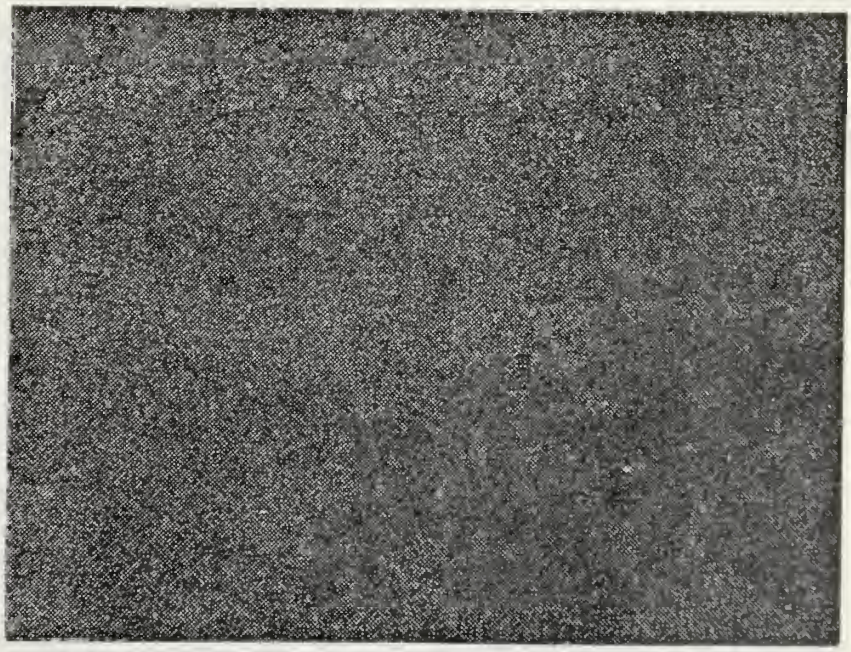

FIGURE 25. K4-6, After one Flop-Rolling with 10% Deformation at 835°C (50X)

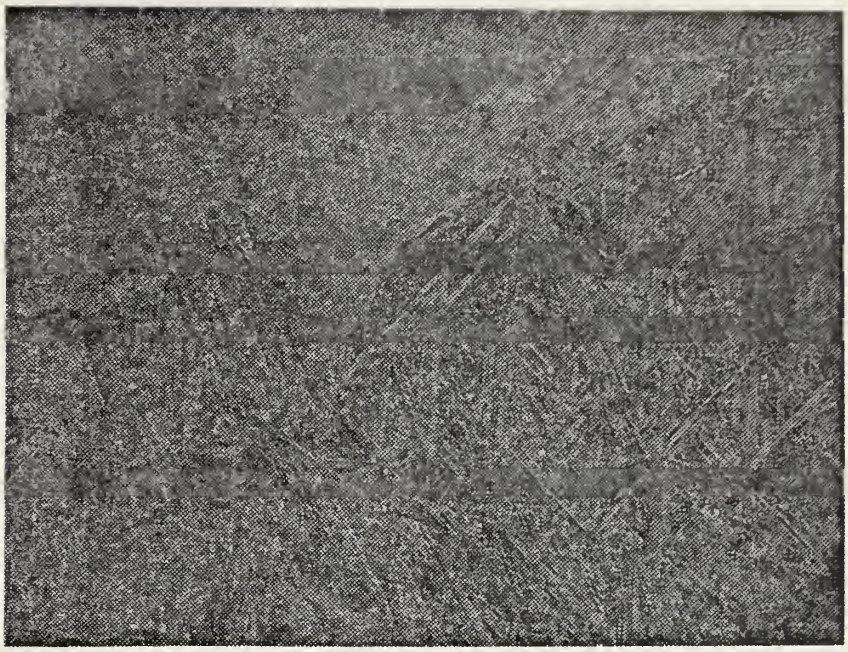

FIGURE 26. K4-6A (h=0.1385"), Annealed at 835°C for 90 Seconds (50X)

FIGURE 27. K4-6B (h=0.1390"), Annelaed at 835°C for 80 Seconds (50X)

FIGURE 28. K4-6C (h=0.1300"), Annealed at 835°C for 70 Seconds (50X)

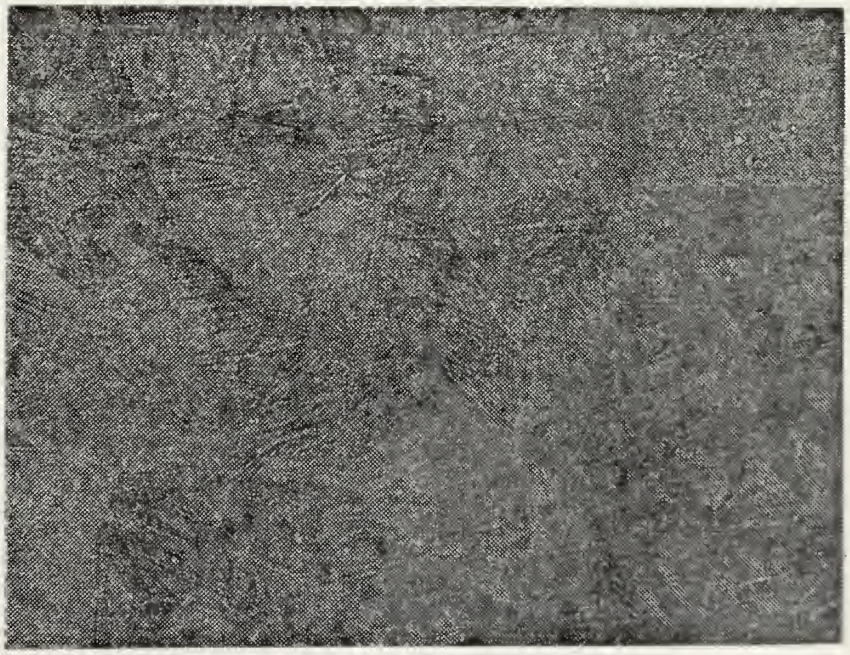

FIGURE 29. K4-6D (h=0.1050"), Annealed at 835°C for 60 Seconds (50X)

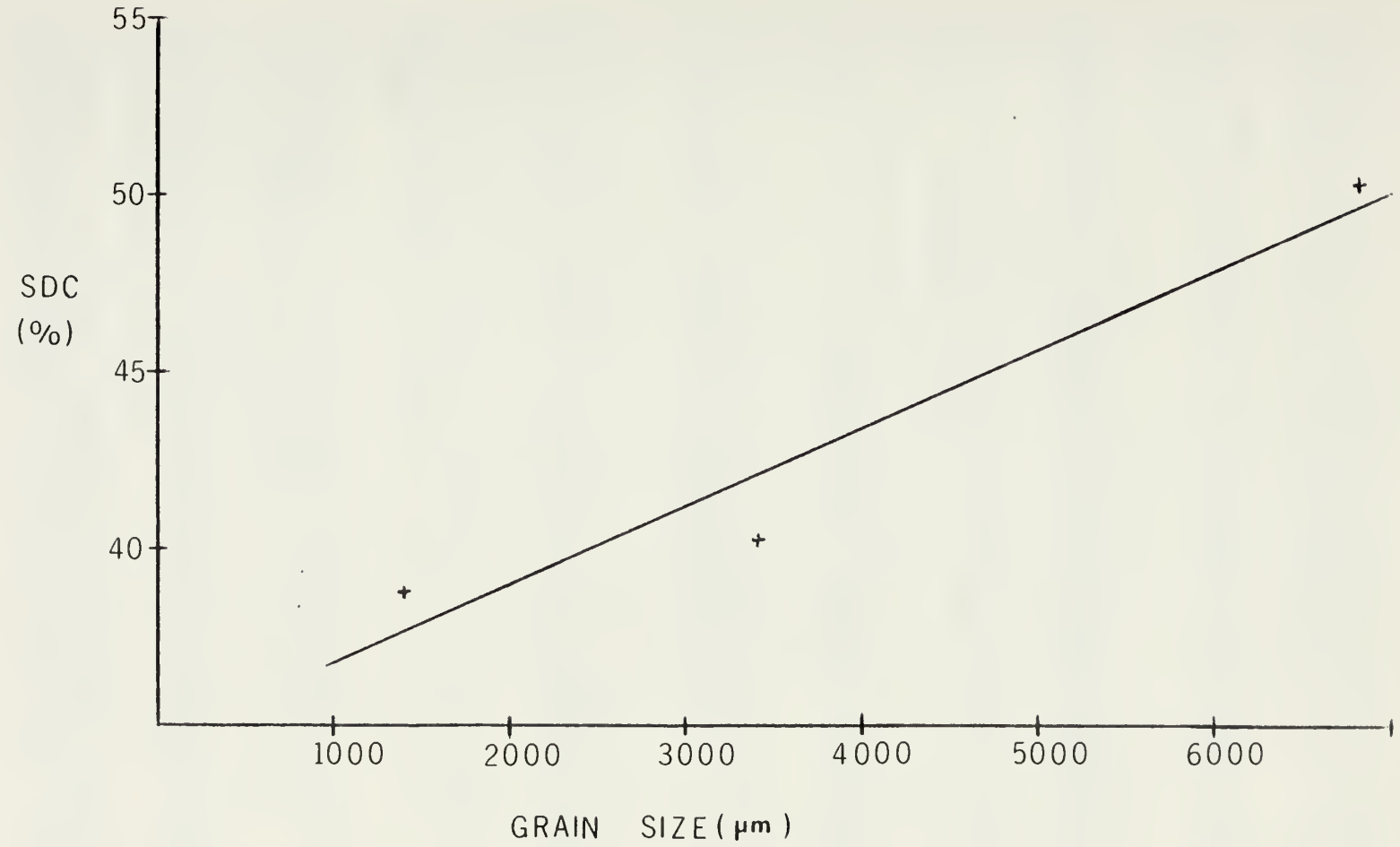


FIGURE 30. Specific Damping Capacity Versus Prior Austerite Grain Size

## LIST OF REFERENCES


1. Langham, John M., A New High Damping Alloy, paper presented at the Pacific Conference of the American Foundrymen's Society, 17th, April 1968.

2. Olin Metals Research Labs, INCRA Project No. 205 Final Report, "Incramute I: Processing and Properties", July 1973.

3. Naval Ship Research and Development Center Report 3236, Development of Improved Alloys for Navy Ship's Propellers, by E. J. Czyryca and M. G. Vassilaros, June 1971.

4. Naval Ship Research and Development Center Report 8-975, Corrosion Behavior of Manganese-Copper-Vanadium Alloy, 16 November 1971.

5. Olin Metals Research Laboratories, Final Report INCRA Project No. 221, Improved High Damping Copper-Base Alloys, by J. C. Fister and S. Shapiro, September 1974.

6. Army Materials and Mechanics Research Center Report 74-53, Noise Abatement and Internal Vibrational Absorption in Potential Structural Materials, by S. A. Kulin, L. Kaufman and P. P. Neshe, August 1974.

7. Department Metaalkunde, Katholieke Universiteit Leuwen Annual Report INCRA Project No. 238, Shape Memory Effect and Internal Damping in Copper-Aluminum and Copper-Zinc Based Alloys, by L. Delaey and others.

8. W. De Jonghe, et al., "Internal Frcition Measurement on Copper-Zinc-based Martensite", pp. 451-466 in Shape Memory Effects in Alloys (J. Perkins, editor), Plenum, 1975.

9. Entiwistle, K. M., chairman, "Discussion: The Physical Metallurgy of Alloys of High Damping Capacity", Journal of the Institute of Metals, v. 93, p. 546-549, 1964-65.

10. Basinski, Z. S. and J. W. Christian, "The Cubic-Tetragonal Transformation in Manganese-Copper Alloys", Journal of the Institute of Metals, v. 80, p. 659-666, 1951-52.

11. Birchon, D. et al., "Mechanism of Energy Dissipation in High Damping Capacity Manganese-Copper Alloys", Metal Science Journal, v. 2, p. 41-46, March 1968.

12. Klier, E. P. and S. M. Grymki, "The Transformations in β-Cu-Al Alloys", Metals Transactions, v. 185, p. 611-620, September 1949.

13. Swann, P. R. and H. Warlimont, "The Electron-Metallography and Crystallography of Copper-Aluminum Martensites", ACTA Metallurgica, v. 11, p. 511-526.

14. H-Warliwont, "Microstructure, Crystal Structure and Mechanical Properties of Martensite Phases in Copper Alloys", pp. 58-67 in Physical Properties of Martensite and Bainite, Special Report No. 93, The Iron and Steel Institute, 1965.

15. Youngblood, F. L., Characterizing and Controlling the Metallurgical Properties of a Cu-Mn Alloy for Ship Silencing Applications, Master's Theis, Naval Postgraduate School, 1975.

16. Goodwin, R. J., "Manganese-Copper Alloys of High Damping Capacity", Metal Science Journal, v. 2, page 121-128, July 1968.

17. Hedley, J. A., "The Mechanism of Damping in Manganese-Copper Alloys", Metal Science Journal, v. 2, page 129-137, July 1968.

18. Hills, N. A., A Study of the Influence of Stream and Temperature on the Damping Capacity of Mn-Cu Alloys for Ship Silencing Applications, Mechanical Engineer Thesis, Naval Postgraduate School, 1964.

19. S. A. Kulin, L. Kaufman, and P. P. Neshe, "Noise Abatement and Internal Vibrational Absorption in Potential Structural Materials", Man Labs Report to Army Materials and Mechanics Research Center, AMMRC-CTR-74-53, August 1974.

20. Jellison, Jane and E. P. Klier, "The Cooling Transformations in the Beta Eutectoid Alloys of the Cu-Al System", Transactions of The Metallurgical Society of AIME, v. 233, p. 1694-1702, September 1965.

21. United States Department of the Interior, Bureau of Mines Report of Investigations 5441, Damping Capacity — Its Measurement and Significance, by J. W. Jensen, p. 4-27, 1959.

22. Guy, A. G. and Hren, John J., Elements of Physical Metallurgy, p. 228-238, Addison-Wesley Publishing Company, 1974.

23. Perkins, Jeff, editor, Shape Memory Effects in Alloys, p. 273-303, Plenum Press, 1975.

24. Naval Ship Systems Command Ship Silencing Program Report NPS-59Ps74061, Materials Approaches to Ship Silencing, by J. Perkins, G. R. Edwards and N. Hills, June 1974.

25. Jensen, J. W. and J. A. Rowland, Jr., "High Damping Alloys", Product Engineering, p. 135-137, May 1956.

INITIAL DISTRIBUTION LIST

| | | No. of Copies |
|---|---|---|
| 1. | Defense Documentation Center<br>Cameron Station<br>Alexandria, Virginia 22314 | 2 |
| 2. | Library, Code 0142<br>Naval Postgraduate School<br>Monterey, California 93940 | 2 |
| 3. | Department Chairman, Code 69<br>Department of Mechanical Engineering<br>Naval Postgraduate School<br>Monterey, California 93940 | 1 |
| 4. | Professor A. J. Perkins, Code 69Ps<br>Department of Mechanical Engineering<br>Naval Postgraduate School<br>Monterey, California 93940 | 3 |
| 5. | LT Edward W. Kelly<br>26 Woodhaven Lane<br>Willing Boro, New Jersey 08046 | 1 |

Thesis
K278
c.1
174742
Kelly
Effects of thermomechanical processing on damping characteristics of martensitic Cu-13.5 w/o an alloy for ship silencing application.